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"CEROXIDE" DEFECT*—WHAT IT IS AND HOW ITS INCIDENCE AND SIZE CAN BE REDUCED

*By Herbert H. Blosjo***

A considerable amount of work relative to a "Ceroxide" type of defect has been done at Minneapolis Electric Steel Castings Company since the 1955 T & O Conference. The results have been quite positive and are definite enough that a theory of Ceroxide formation, which seems to fit the facts, is proposed for consideration.

A definition of Ceroxide as it is now understood, and which refers to the originally named defect that has become the bane of steel foundrymen, is now apropos. A Ceroxide defect, which is illustrated in Figure 1, is a shallow depression that is usually found on the cope surface of a steel casting, occasionally on a vertical surface, and is caused by the formation of a low fusing, nonmetallic substance during the time the mold is poured; it does not originate from the ladle lining. Defects, quite similar to Ceroxides, can be obtained from floor ladle conditions, but these defects are a different problem and will not enter into the following discussion.

Previous work indicated that this slaggy material was primarily a result of the fusion of bentonite by the hot metal. Recent work indicates that this conclusion is erroneous and the work of Koch[1] and Sanders[2] has substantiated the latter opinion in some respects.

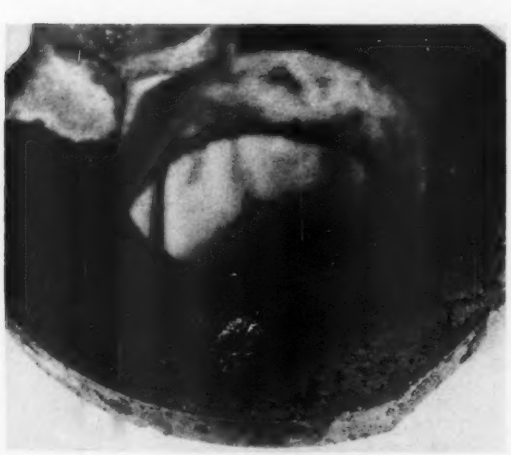

Figure 1—An example of a Ceroxide. Some of the Ceroxide material remains in the defect.

The following numbered tests were run at least in duplicate and in most cases in relatively large numbers:

1. Green sand molds made from heap sand and poured without drying.
2. Green sand molds made from heap sand and sprayed fairly heavy with Top Bond "B" all over the mold cavity, riser and gate surfaces, closed immediately and poured without drying.
3. Green sand molds made from heaps and rebonded to high green strength with western bentonite.
4. Molds made from new sand facing and poured without drying.
5. Molds made using new sand facing. Molds were completely dried with infra-red dryers, closed and poured.
6. Molds made from new sand facing with an addition of 3 percent organic material. These molds were completely dried with infra-red dryers, sprayed with Top Bond "B", closed and poured.
7. Baked oil sand molds with good surface hardness.
8. Baked oil sand molds with four percent western bentonite in the core mix. Surface of core was soft due to the bentonite absorbing much of the oil.

The pattern used for these molds was the SFSA six-inch square by 5-inch deep block. A 3-inch diameter head was centrally located on the cope side of the casting. Some of the tests were repeated on production castings; in fact, production procedures are very closely geared to the results obtained in these tests. The SFSA cube casting is a very severe test. It has been found that if the sands and procedures produce a good SFSA cube casting, then the same sands and procedure will give excellent results in production.

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* "Ceroxide will use in the Journal of Steel Castings Research when referring to the shallow smooth defects often found on the cope surfaces of steel castings and will be used until a better term is suggested.

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The castings obtained from the above molds were graded in two respects, i.e., the amount of Ceroxide and the amount of angular defects (loose unfused sand). The test numbers, referring to those shown above, are listed in the order of increasing severity of defect observed on the casting produced. The test producing the best casting, therefore, is listed first.

Ceroxides	Angular Defects
Test No. 7	Test No. 7
Test No. 8	
Test No. 6	
Test No. 5	
Test No. 4	
Test No. 2	Test No. 1
Test No. 3	Test No. 8
Test No. 1	Test No. 6

The molds were poured with low carbon acid electric steel having an addition of 2.5 pounds of No. 2 aluminum added per ton of melt. The aluminum was added as shot to the stream while tapping into a tea pot ladle. No other deoxidizer was used. The tea pot ladle was lined with zircon. Metal was transferred to a silica lined 300 pound monorail ladle with a deep skimmer. It was found that molds generating a reducing atmosphere prevented or reduced the incidence and quantity of Ceroxides. The greater the reducing character of the mold atmosphere, the fewer the defects. Molds with a reducing atmosphere but having a friable surface did not show any Ceroxide defects. In their place were small angular defects containing eroded sand from the mold.

The amount of Ceroxides forming in molds of similar facing, but with some of them poured green and others poured after drying, was very nearly the same. However, the dried molds on the average showed a little less of the Ceroxide material. This indicates very strongly that the cause of the Ceroxide is not moisture itself but oxygen. The oxygen in the moisture only aggravates the existing oxidizing condition in the mold.

A number of foundrymen believe that Ceroxide formation takes place only in small molds. This is not true, as they are found in larger molds. In larger castings, the Ceroxide remains more fluid and the rising metal heats the sand so that the fluid material flows into the interstices of the sand. The Ceroxide material actually penetrates the mold surface instead of displacing the metal of the casting. When shaking out large castings, the sand will peel off to a nice shiny surface on most of the cope, but sand will adhere in some places. These places are where the Ceroxide material contacted the sand. If the sand and casting are examined at this place, the following may be observed. The sand will have a fairly large void which is indicative of the corrosiveness of the fluid material. Actually it dissolves some of the sand on the surface of the mold. The casting will usually have a spangled appearance where this fluid material contacted the casting but the contour of the casting will conform to the original mold. After annealing and blasting the casting it is impossible to tell that there had been a Ceroxide formed. Since the Ceroxide formation in large molds does not form a defect in the casting, many foundrymen believe that Ceroxides do not form in large molds.

Ceroxide material has been analyzed by a few people. Although their method of sampling is not specified, it is believed that the samples were taken from the casting—a Ceroxide sample taken from a casting after the defect has come into contact with the sand mold is not satisfactory. Such a sample does not represent the particular material which lodged at that point but is a solution of the mold surface in the fluid material which did lodge there.

Many Ceroxide defects have been observed in castings. Almost without exception there will be sand grains visible in the slaggy material. Actually the Ceroxide material does pick up sand grains while pouring but the material is so corrosive that it dissolves these sand grains almost instantaneously.

The sand grains found in the Ceroxide on castings are sand grains which were pulled out of the face of the mold. In several instances, during the pouring of a mold, Ceroxide samples were taken by means of a steel rod inserted through the riser opening, before the metal contacted the cope surface. None of these had any indication of sand grains in them, although without question some eroded grains were in solution.

A Ceroxide sample was sent to an analytical laboratory for semi-quantitative spectrographic analysis. This sample was a composite of a large number of small samples taken out of molds while pouring. The following is the analysis:

SAMPLE No. 1	Percent
Silicon Dioxide	60.0
Ferrous Oxide	15.0
Aluminum Dioxide	13.0
Manganous Oxide	11.0
Calcium Oxide	0.5
Titanium Dioxide	0.3
Magnesium Oxide	0.1
Chromic Oxide	0.1
Nickel Monoxide	0.02
Zinc Oxide	0.01
Cupric Oxide	0.01
Stannous Oxide	0.005
Molybdenum Oxide	0.001

Sanders reports the analysis of Ceroxide taken from sand and graphite molds as follows:

SAMPLE No. 2

	Sand Mold Percent	Graphite Mold Percent
Silicon Dioxide	84.0	43.0
Ferrous Oxide	4 to 5	2.0
Aluminum Trioxide	5	18.0
Manganous Oxide	4 to 6	14.0
Calcium Oxide	1	18.0

First, it will be noted that the silicon dioxide content of Sanders' sample was much higher than in Sample No. 1. This is perhaps due to one of two conditions. If the sample was taken out of the molds before coming in contact with the mold face, it would indicate that the mold was of poor quality and that a large amount of sand had eroded and dissolved in the other fluid oxides. This would indeed require a mold of very poor surface quality, as will be shown later. What more likely happened is that the samples were taken out of the casting and were diluted with silica from the mold surface at that point.

There is very close similarity in the above two analyses of Ceroxides except for the silicon dioxide content. This is important and will be discussed later. If the silica content of Sample No. 1 was nearly quadrupled, the analysis would be practically the same as found by Sanders. This is shown in Table 1 by adding 170 grams of silica to 100 grams of the Ceroxide and recalculating the analysis.

This would indicate that the corrosive oxides which attacked the silica of the mold were of about the same analysis in both cases. The above data lends support to a new theory of Ceroxide formation which is defined below.

The slaggy material in a mold which causes the Ceroxide defect is the result of the oxidation of a number of elements in the steel due to an oxidizing atmosphere in the mold. The resulting corrosive constituent reacts with the molding material, particularly eroded mold particles, to form a viscous material which lodges in the casting.

In sand molds the Ceroxide analysis will vary due to the condition of the mold and the method of gating. Molds which erode easily will have Ceroxides of higher silica content and, therefore, under these circumstances the Ceroxide volume is not only increased but also the viscosity is increased. This results in a larger and deeper cavity in the casting. It is believed that large molds normally have less erosion and better conditions for oxidation than is usually found in small molds. This would result in a Ceroxide of low viscosity. Also, the radiant heat from the larger volume of metal heats the cope sand to such a degree that the fluid material penetrates the mold surface with the result that no defect is found in the casting.

Sanders also found a Ceroxide material in graphite molds. The analysis indicated a much lower silica content, and all the other oxides were increased except the iron oxide. The calcium oxide was particularly high, this indicates that the metal was deoxidized with a calcium alloy. The ferrous oxide was very low indicating considerably less oxidizing atmosphere in this mold than in a sand mold. The silicon dioxide is lower as it cannot be diluted by erosion of the mold. The high alumina is no doubt due to the use of aluminum as well as calcium in the final deoxidation of the melt.

Exception is taken to Sanders' statement that this Ceroxide material is slag sucked in through the gate. No doubt some slag may be sucked into the mold which may increase the volume of Ceroxide, but basically, the Ceroxide is oxides formed in the mold due to an oxidizing atmosphere. Poor gating adds to the volume of Ceroxides in at least two ways. Turbulent gating gives greater contact between incoming metal and mold atmosphere resulting in the formation of larger amounts of corrosive oxides. Turbulent gating also causes more erosion of the mold which results in a greater amount of Ceroxides of higher viscosity. The higher viscosity results in deeper defects in the casting. Data substantiates Sanders' statement that calcium-manganese-silicon and aluminum contribute to Ceroxide formation. In fact, any element which oxidizes easily would contribute to Ceroxides.

TABLE 1—Recalculation of Ceroxide Composition

	MESCO Analysis		Calculated Analysis Percent	Sander's Analysis Percent
Silicon Dioxide	60.0 g + 170 g =	230.0 g	85.0	84.0
Ferrous Oxide	15.0	15.0	5.5	4 to 5
Aluminum Trioxide	13.0	13.0	4.8	5.0
Manganous Oxide	11.0	11.0	4.1	4 to 6
Other Oxides	1.0	1.0	0.4	1.

Koch believes Ceroxides are formed by free oxygen attacking eroded sand from the mold. The free oxygen coming from the disassociation of water vapor in contact with the liquid steel. Certainly water vapor aggrevates the Ceroxide condition but it is only a contributing factor. The water vapor is disassociated into hydrogen and oxygen and both are absorbed by the liquid steel. The oxygen, however, reacts with the aluminum, manganese, silicon and other elements to form a corrosive constituent which in turn fluxes eroded sand from the mold. Oxygen from the mold atmosphere will create Ceroxides in the same way.

The incidence and size of Ceroxide can be reduced by the following means:

1. Ceroxides do not form in baked sand cores because of a reducing mold atmosphere.
2. In molds bonded with bentonite and cereal the following items decrease ceroxide formation.
   a. The use of high hot strength bentonites.
   b. The use of new facing.
   c. The molds should be rammed to high hardness.
   d. Skin dried molds are slightly better than green molds.
   e. Gate to reduce turbulence.
   f. Spray mold cavities and gates with a solution of high volatile solvent and plastic. (This is one of the most effective means.)

References

1. Norman F. Koch, "A Study of the Snotter Cope Defect and Erode Sand," STEEL FOUNDRY FACTS, July, 1956.
2. Clyde A. Sanders, "One Answer to the Snotter Problem," MODERN CASTINGS, May, 1956.

CORRELATION OF THE PROPERTIES OF MOLDING SANDS WITH HOT TEARING SUSECEPTIBILITY*

A recent study conducted at the Naval Research Laboratory by N. C. Howells, R. E. Morey and H. F. Bishop on the Properties of Molding Sands Under Conditions of Gradient Heating revealed some interesting results on the hot tearing susceptibility of various sand mixes. It was observed that castings made with cores of high density molding sand and with sand containing 10 percent silica flour, both of which showed high strengths and elastic moduli in the gradient heating tests, developed sever bore cracks. No bore cracks were observed in castings containing cores made with either the low density molding sand or with sand containing one percent vacuum dried ammonium nitrate, both of which showed low strengths and moduli in the gradient tests at the 6-minute exposure time.

On the other hand, cores made with sands bonded with 1 - 2 percent oil, which was relatively weak after 6 minutes, and with sands bonded with 2 percent rosin, which had the lowest strength and modulus of all the mixtures tested, developed bore cracks. On the basis of these tests, susceptibility to bore cracking cannot be predicted from the hot strength of the core sand alone.

A general correlation of hot tearing severity with the dry density of the core was found to exist. This observation supports the results of previous investigations. It appears that if hot tears are to be eliminated, it is of primary importance that the molding sand be compressed under the contraction stresses of the solidifying casting.

The rosin-bonded sand, while it has a low strength and modulus, has very high flowability and is easily rammed into a dense core. As a result its degree of compaction cannot be signifantly increased during the hot-tearing interval; casting contraction is thus restrained to cause hot tears. The green sand containing the solid ammonium nitrate additive had a very low flowability.

The ammonium nitrate liquified upon exposure to heat and caused the sand to become mushy and flowable. Thus, under actual molding and casting conditions, the mushy layer of sand in the mold could be compacted to permit the relief of casting strains which would otherwise lead to hot tearing. On the basis of these tests, the degree of compaction, measured as dry density, appeared to be a prime factor for hot tearing susceptibility rather than any particular level of high-temperature strength.

* Abstracted from NRL Report 4857, "Properties of Molding Sands Under Conditions of Gradient Heating", published by the Naval Research Laboratory on October 29, 1956.

CALCIUM IN STEEL CASTINGS

*D. C. Hilty and F. J. Shartsleeve**

A summary treatment of the chapter on "Calcium in Iron and Steel" from the 4th Monograph of the Alloys of Iron Monograph Series "Columbian, Tantalum, Zirconium, Boron and Calcium in Iron and Steel"

Calcium Alloys and Compounds

Although calcium is used rather extensively as an addition agent in the production of iron and steel, it is employed more frequently in the form of alloys or compounds of calcium than as elemental calcium. A great many calcium-bearing complex addition agents have been recommended for various applications. However, calcium-silicon, calcium-manganese-silicon, and calcium carbide are used most extensively. In these forms calcium is used to deoxidize, desulfurize, and degasify steel and cast iron, to control the type and distribution of nonmetallic inclusions in steel, and to graphitize and promote a uniform microstructure in gray iron.

A wide variety of alloys has been prepared containing between 15 and 30 percent calcium and between 40 and 70 percent silicon, the balance being mainly iron. However, the composition of calcium-silicon generally used as an addition agent in steelmaking is in the range of 28-35 percent calcium, 60-65 percent silicon, and less than 6 percent iron. Calcium-manganese-silicon generally contains 16-20 percent calcium, 16-17 percent manganese, and 53-59 percent silicon.

It is now generally agreed that calcium, introduced by the usual metallurgical techniques, is virtually insoluble in liquid and solid steel. The small amounts of calcium occasionally reported in steel are believed to be present in the nonmetallic inclusions. Such inclusions have been chemically removed from steel, and on a qualitative basis have been shown to contain calcium.[1]

Deoxidizing Power of Calcium Alloys

The results of a great many plant-scale tests and laboratory investigations indicate that calcium added under carefully controlled conditions is an effective deoxidizing agent. The opinions of many investigators concur that calcium-silicon is an effective deoxidizing agent, and that the deoxidation product rises to the surface of the bath, leaving no harmful nonmetallic inclusions in the steel.

The extent to which the oxygen content of steel may be reduced by calcium may be estimated from thermodynamic data. When a deoxidizing agent such as calcium is added to steel, it reacts with the dissolved oxygen until the activity of the dissolved oxygen is reduced to a definite value described by the equilibrium constant.

On theoretical grounds, it appears that the oxygen content of steel might be reduced to extremely low values by the addition of calcium.

In 1934, Kroll[2] stated, "The ability of calcium as a scavenger for steel is rather doubtful. Calcium does not dissolve in iron, so that the reaction takes place only superficially. Calcium boils at 2622 degrees F so that in molten steel it can exist only as a vapor." Other investigators have felt that, because of its marked tendency to escape into the atmosphere, calcium additions effect little or no deoxidation under normal steelmaking conditions.

In 1938, Sims and Dahle[3] attempted to deoxidize cast steel with calcium-silicon additions of 0.05-0.15 percent. The nonmetallic inclusions in the solid steel had practically the same appearance as those observed in silicon-killed steel. The investigators considered this result as evidence that calcium additions produce no deoxidation. However, they pointed out that the calcium may have failed to enter the steel because it was added to the bare surface of the bath and no effort was made to hold the alloys under the surface of the molten steel. In 1949, Sims, Saller, and Boulger[4] reported that the deoxidizing power of calcium is no stronger than that of silicon, and that calcium additions are unable to reduce the FeO content of steel sufficiently to form Type II intergranular sulfides.

The diverse opinions and apparent contradictory experimental evidence concerning the utility of calcium, may be explained by the fact that consistent results can be obtained only under carefully controlled operating conditions and that variations in tapping temperatures, timing of additions, and tapping technique may have a pronounced effect on the utility of a given addition of calcium.

There can be little doubt that calcium-bearing addition agents when added in a suitable manner are effective in reducing the oxygen content of steel. However, this effect is probably the result of an influence of calcium on the kinetics of the deoxidation reaction, and true equilibrium between calcium and oxygen dissolved in the metal is not attained.

It has been reported by several investigators that calcium-silicon additions are effective in reducing the sulfur content of steel. In 1934, Kroll[2] questioned the usefulness of calcium and cited instances where calcium-silicon alloys were employed

*Metals Research Laboratories, Electro Metallurgical Co., Division of Union Carbide and Carbon Corporation.

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for the desulfurization of steel without success, which again illustrates the necessity of extreme care.

Evidence that calcium-manganese-silicon additions are effective in reducing the sulfur content of cast steel was presented by Wyman.[5] In the course of a deoxidation study, he added calcium-manganese-silicon to cast steel containing 0.30 percent carbon to introduce calcium contents of 0.05, 0.10, 0.15, 0.20, 0.25, and 0.30 percent by weight. Sufficient aluminum was added to each heat to give a final aluminum content of 0.085 percent. The effects of increasing amounts of calcium on the sulfur content of steel are shown in Figure 1. In the plot it may be seen that the heaviest additions reduced the sulfur content from 0.043 to 0.014 percent. Wyman also reported a drop in the phosphorus content from 0.040 to 0.033 percent and a marked increase in the strength and ductility of the steel. The increase in tensile strength and yield strength was attributed to the increased manganese and silicon contents. The increase in ductility was believed to be produced by the decrease in the sulfur content occasioned by the desulfurizing effects of calcium. However, in Wyman's experiment, the increasing calcium adidtions were necessarily accompanied by increasing silicon contents, so that strongly reducing conditions were not only maintained but intensified.

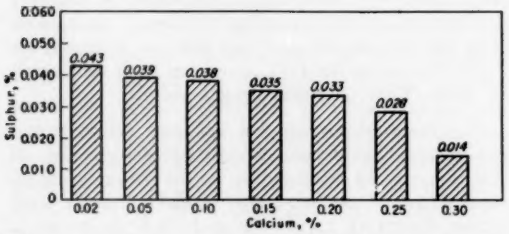

Figure 1—Effects of calcium additions as calcium-manganese-silicon on the sulphur content of steel. Wyman[5]

A number of investigators have shown that calcium carbide is an effective desulfurizing agent. The results of investigations carried out by Carter[6] show that injection of calcium carbide by means of a carrier gas can also be employed to remove sulfur from both acid and basic steel. The efficiency of desulfurization, however, is low. The quantity of carbide required varied from 35 to over 100 pounds per pound of sulfur removed, depending both on the carbon content of the metal and on the initial sulfur content. The efficiency was highest when the initial sulfur and the carbon content of the steel were highest.

Johannsen[7] found that the melting point of calcium carbide was lowered by the addition of fluorspar and patented a mixture of calcium carbide, fluorspar, and cryolite to be used for the deoxidation of finished steel outside the furnace.

It appears that the sulfur content of steel can be reduced by the addition of calcium. However, special techniques or devices must be employed for its introduction into the molten metal if large reductions in sulfur content are to be effected. Moreover, calcium seems to be considerably more effective as a desulfurizer for high-carbon iron than for steel. Since calcium has a greater affinity for oxygen than for sulfur, strongly reducing conditions are necessary for effective desulfurization, and these conditions are attained more readily in high-carbon iron than in steel.

Effects of Calcium on Nonmetallic Inclusions

Although calcium additions have the ability to reduce the oxygen content and to some extent the sulfur content of steel, they are not widely used specifically for this purpose. Deoxidation is generally accomplished by use of silicon, aluminum, or other agents, and calcium is added to the deoxidized steel to control the type and distribution of nonmetallic inclusions. It is for this purpose that calcium additions, principally in the form of calcium-silicon and calcium-manganese-silicon, find their major application in the production of steel. It has been shown by several investigators that calcium additions of 0.05-0.1 percent tend to eliminate the intergranular eutectic type of inclusions so detrimental in steel castings.

Sims and Dahle[3] classified three standard inclusion types that result from deoxidation with various amounts of aluminum. They may be described as below.

Inclusion Type I consists of randomly distributed globular silicates and sulfides of a wide range of sizes. This type of inclusion is obtained in silicon-killed steel when no aluminum is used. It is characteristic of cast steel with high ductility. The advantages associated with aluminum deoxidation, such as degasification and absence of pinhole porosity, cannot be easily realized if Type I inclusions are to be obtained.

Inclusion Type II consists of sulfides distributed in a eutectic pattern in the primary grain boundaries with occasional clusters of alumina. The amount of aluminum required to produce this type of inclusion varies with steelmaking practice, but it is of the order of 0.025 percent. Type II inclusions are associated with low ductility.

Inclusion Type III is produced by relatively high additions of aluminum (0.10-0.20 percent). The inclusions again consist of grain-boundary sulfides and alumina, but the sulfides are larger and farther apart. They do not form a continuous zone of weakness, and steel containing this inclusion

type is more ductile than steel containing Type II inclusions.

The average values for ductility of Grade-B steel castings containing the three types of inclusions are shown in Table 1.

TABLE 1—Average Values for Ductility of Grade-B Steel Castings[3]

Typical inclusions	Type	Approximate aluminum addition, %	Elongation in 2 in.. %	Reduction of area, %
Silicates	I	None	30	49
Eutectic	II	0.05	23	35
Alumina	III	0.10	27	42

Sims and Lillieqvist[8] had previously suggested that the formation of fine intergranular (Type II) inclusions is detrimental to the ductility of steel and thus limits the amount of strong deoxidizers which may be used. However, as stated above, Sims and Dahle found that treatment with larger amounts of aluminum produces Type III inclusions and consequently tends to eliminate the harmful eutectic inclusions.

In discussion of Sims and Dahle's paper[3], Grotts recommended a deoxidation treatment in which high aluminum additions are supplemented with calcium-silicon. Average results under commercial test conditions in which a steel containing 0.23-0.28 percent C, 0.65-0.75 percent Mn, 0.28-0.33 Si, and 0.04 percent S and P, each, was treated with 0.135-0.175 percent aluminum were reported as shown in Table 2. Grotts concluded from these data that calcium-silicon used in conjunction with aluminum is very beneficial.

Similar results were reported in the same year by Gagnebin[9] who proposed a new deoxidation method which, he claimed, combined very good ductility with very low oxygen content. The method consists of producing steel in the regular manner, oxidizing it, and then finishing with calcium (0.10-0.20 percent) and aluminum (0.05-0.10 percent). Gagnebin stated that the addition of a strong sulfide-forming element such as calcium, in the presence of sufficient oxide content, causes random dispersion of the inclusions even though the oxide content may be markedly reduced by the addition of a strong deoxidizer such as aluminum.

TABLE 2—Properties of Steel after Supplementing the Aluminum Used in Deoxidation with about 0.155% Calcium Silicide*

Deoxidation	Tensile strength, psi	Yield point, psi	Elongation in 2 in., %	Reduction of area, %
Plain Al	72,000	43,000	29.2	46.2
Al+CaSi	74,000	45,350	31.2	54.2

* Grotts discussion of Sims and Dahle[3].

In 1940, Crafts, Egan, and Forgeng[10] carried out a rather extensive investigation of the mechanism of inclusion formation in steel castings. They deoxidized the steel castings with graduated amounts of aluminum and studied the effects of supplementary calcium additions on the inclusions in highly deoxidized steel. Calcium alloys were added either in the form of loose lumps or wrapped in a light perforated sheet-steel capsule and plunged into the metal. Typical properties of medium-carbon steel castings made according to different types of deoxidation treatment are shown in Table 3, where it may be seen that calcium added in conjunction with aluminum caused a marked improvement in the tensile ductility of the steel.

Crafts, Egan, and Forgeng identified five types of inclusions associated with various deoxidation practices, as described in Table 4. Of these five, three types—silicates, eutectic, and alumina—correspond to inclusion Types I, II, and III as described by Sims and Dahle. The galaxy inclusions represent an intermediate, undesirable condition. The peritectic type, which was produced by a supplementary addition of 0.05 percent calcium, is characteristic of good ductility and consists chiefly of alumina, duplex oxide-sulfide inclusions, and round sulfides.

With regard to the effects of calcium on nonmetallic inclusions Crafts, Egan, and Forgeng stated that calcium-bearing alloys added to steel which had received an addition of 0.10-0.20 percent aluminum, titanium, or zirconium produced duplex inclusions with an angular idiomorphous core surrounded by sulfide-like material. In varying degrees, depending upon the amount and efficiency of the calcium addition, the duplex inclusions tend to become

TABLE 3—The Effect of Calcium Deoxidation on the Mechanical Properties of Steel Castings [10]

Deoxidizing addition		Tensile strength, psi	Yield point, psi	Elongation in 2in., %	Reduction of area %
Al	Ca				
0.10	—	75,750	50,000	27.5	30.4
0.10	0.10	82,000	49,500	30.0	50.0

TABLE 4—Types of Inclusions [10]

Type	Aluminum addition, %	Ductility	Nonmetallic inclusion
Silicates	0.025	Very good	Glassy silicates, round sulfides
Eutectic	0.05	Poor	Round sulfides, and intergranular sulfide films
Galaxy	0.075	Poor	Alumina, round sulfides, and films or oxide galaxies
Alumina	0.10	Fairly good	Alumina, round sulfides
Peritectic	0.10	Good	Alumina, duplex, round sulfides

more numerous and the oxide galaxies tend to disappear. It was suggested that the duplex inclusions are formed by a peritectic reaction in which the oxide core does not have an opportunity to be entirely converted to oxysulfide compound. However, as subsequently explained by Crafts and Hilty[11] it is more probable that the inclusions result from an oxide-sulfide eutectic which solidifies during freezing of the metal. Crafts, Egan, and Forgeng believed that the ductility of steels of the peritectic type is superior to that of the alumina type because of the elimination of galaxy inclusions. The ductility of steel containing the peritectic type of inclusions approaches closely that of steels containing the silicate type. Steel castings containing the alumina and peritectic types of inclusions are relatively resistant to hot tearing at the oxidized surface owing to the formation of separated oxide particles rather than continuous intergranular films.

A great many plant-scale tests have substantiated the fact that supplementary calcium additions, as calcium-silicon or calcium-manganese-silicon, have a beneficial effect on the inclusions in deoxidized steel.

In 1947, in reviewing deoxidation practices in basic steel foundries, Lillieqvist[12] stated, "It has been our experience that the only deoxidizer that does completely eliminate pinhole porosity is aluminum. The use of calcium-silicon or calcium-manganese-silicon additions with aluminum is fairly common. These extra additions seem to provide more consistent inclusions. The exact mechanism is still rather uncertain as to whether the calcium acts through the formation of stable sulfides or through a deoxidizing effect which thus leaves more residual aluminum. . . ."

Calcium in Cast Steel

As may be inferred from the preceding, calcium, mainly in the form of calcium-silicon or calcium-manganese-silicon, is extensively employed by the steel castings industry for the inclusion control necessary for the maintenance of uniformly high ductility. Acid open-hearth, acid electric furnace, and basic electric furnace operators alike use final deoxidation treatments incorporating calcium, with results entirely similar to the experimental observations described in the foregoing. It has been established by the castings producers that deoxidation with aluminum is desirable for the control of pinhole porosity in castings. In addition, some aluminum may be added for grain-size control. As pointed out by Crafts[13] however, acid steels tend to be high in sulfur and low in oxygen, so that sulfide inclusions predominate. With aluminum deoxidation, segregated oxide clusters or galaxies are formed, which perhaps are not so harmful to the ductility of castings as intergranular sulfides but are none the less objectionable. In both types of steel it is common practice to employ calcium-silicon or calcium-manganese-silicon to minimize the undesirable effects of aluminum.

Although the specific deoxidation practice may vary somewhat from shop to shop with the type of steel being cast, it generally consists of a ladle addition of calcium-manganese-silicon or calcium-silicon accompanied by another strong deoxidizer such as aluminum or zirconium. The additions are usually made to the tapping or "bull" ladle, but some foundries, when pouring a large number of small castings from one heat, may employ supplementary deoxidizing additions, including calcium, in the shank ladles. The usual addition consists of 3 or 4 pounds of calcium alloy per ton of steel and 2-2.5 pounds of aluminum. In specific cases, more or less calcium may prove desirable. For carbon and low-alloy steel castings, calcium-manganese-silicon is normally preferred to calcium-silicon or calcium metal. The reason for this is not absolutely clear, but it is probably associated with temperature. Low-carbon steel and steel casting heats are tapped at relatively high temperatures, and the more concentrated and highly reactive additions such as calcium-silicon or calcium metal presumably react too vigorously to be fully effective. In reviewing the use of calcium in cast steel, Crafts[13] stated, ". . . calcium-manganese-silicon has been found to be more effective in castings and

low-carbon steels while calcium-silicon or calcium metal is used on higher carbon and special steels . . . As little as 2 pounds per ton of calcium-silicon has been found effective in some cases. Calcium-manganese-silicon on a pound-for-pound basis is equally effective in spite of a lower calcium content and is preferred for castings and low-carbon steels. The effective amount may be higher for some steels, and 4 pounds per ton of calcium alloy is commonly used. For special products, much larger additions—for example, 8 pounds per ton of calcium-silicon and additions of ½ pound of calcium metal—have been found effective in producing clean steel."

Another effect of calcium of considerable interest to the steel castings industry is its so-called tendency to improve the fluidity of steel. The mechanism by which this improvement occurs is obscure but it is thought that the increased fluidity is the result of a decrease in surface tension associated with the presence of calcium. Regardless of the mechanism, however, this effect of calcium is apparently established particularly in connection with aluminum-killed and alloy steel castings.

Another interesting application of calcium and calcium alloys in carbon and low-alloy steels is the use of calcium for degasifying the molten bath. The violent agitation resulting from submersion of calcium and calcium-silicon in molten steel is considered to be effective in reducing the hydrogen content of the metal. Hydrogen is a continual nuisance in many high-quality engineering alloy steels such as SAE 4300 and 4600 series, and some steelmakers plunge calcium or calcium alloys into the bath in the furnace to minimize the hydrogen content of their steel.

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LOW TEMPERATURE PROPERTIES OF STEEL CASTINGS—CARBON STEELS

By G. M. Michie and W. J. Jackson**

These studies were undertaken in response to a demand for data on the low temperature properties of materials called upon to operate at sub-zero temperatures. A review of published data had shown that most of the available information on the low temperature properties of steel castings was of American origin. In view of differences between American and British steelmaking practice it was considered that the steel foundry industry in the United Kingdom had little reliable data upon which to undertake the commercial risk associated with the production of castings called upon to meet low temperature test requirements of American and British specifications in this field.

Tests were carried out on two grades of carbon steel, namely 0.15-0.20 percent carbon and 0.30-0.35 percent carbon, conforming to B.S. 592:1950, Grades A and B respectively.

Provision of Test Material

The cast steel coupons were supplied by three steel foundries from basic electric steel. Each foundry supplied coupons from three separate heats to check on the reproducibility of the steel. Particulars of the steelmaking and deoxidation practices employed are recorded in Table 1. Chemical analyses of each heat are recorded in Table 2.

Heat Treatment of Test Material

Heat treatment particulars of the coupons are given in Table 3. The third column of this Table shows the treatment applied in accordance with the suppliers' recommendations, and it will be noted that repeat treatments were carried out in the case of the steels from heats C and E in order to investigate whether there was any variation in impact properties in the normalized condition.

Preparation of Test Pieces

Charpy V-notch impact test pieces were prepared from material in each of the conditions of heat treatment detailed in Table 3, as well as from material in the as-cast condition. All the notches were machined in a direction parallel to the nearest

* The British Steel Castings Research Association, Sheffield, England

TABLE I—Manufacturing Particulars of Steels Supplied*

Heat No.	Finishing Practice De-oxidation and		Coupon Type
0.15 - 0.20% CARBON			
A	Si-Mn	15 lb/ton	Keel Block
	Fe-Mn		
	Fe-Si	16 lb/ton	
	Al	2½ lb/ton to ladle	
B	Fe-Mn		Clover Leaf
	Fe-Si		
	Al	2½ lb/ton to ladle	
C	Al	2½ lb/ton to ladle	Clover Leaf
0.30 - 0.35% CARBON			
D	Fe-Mn		Clover Leaf
	Fe-Si		
	Al	2½ lb/ton to ladle	
E	Al	2½ lb/ton to ladle	Clover Leaf

* All steels made by Basic Electric Process.

casting surface (i.e. in general, at right angles in the columnar crystal growth).

V-notch energy transition curves permitted an assessment to be made of the temperatures likely to be associated with a 15 foot-pound keyhole energy. Approximately three keyhole notched specimens from each steel were then broken at temperatures in the neighborhood of these assessed values, the exact temperature associated with 15 foot-pound being in each case subsequently interpolated from the results obtained.

The temperature that will produce 15 foot pounds in the normalized condition for each of the heats produced is given in Table 2.

Results of Impact Tests

The temperatures associated with the 15 foot-pound V-notch fracture energies, tested in the conditions of heat treatment recommended by the suppliers, and the influence of alternative heat treatments are summarized in Table 4.

In relation to the influence of test piece position, it was found that such differences as were obtained among test pieces removed from the same block, and tested at the same temperature, could not be consistently related to their original position in the test casting. To all intents and purposes, therefore, the material used was considered to have been reasonably homogeneous.

TABLE II—Chemical Analyses and Temperature for 15 Foot Pound Fracture

Heat No.	Chemical Analysis C%	Si%	Mn%	S%	P%	Temperature* For 15 ft. lb.
0.15 - 0.20% CARBON						
A						
1	0.15	0.38	0.75	0.032	0.032	−58°F
2	0.20	0.38	0.63	0.034	0.019	−13
3	0.195	0.48	0.83	0.025	0.043	−22
B						
1	0.16	0.38	0.91	0.013	0.019	−103
2	0.17	0.45	0.87	0.016	0.018	−67
3	0.18	0.49	0.95	0.019	0.020	−67
C						
1	0.19	0.47	0.67	0.015	0.037	−49
2	0.16	0.38	0.60	0.018	0.036	−49
3	0.17	0.44	0.77	0.018	0.027	−49
0.30 - 0.35% CARBON						
D						
1	0.30	0.30	0.81	0.026	0.015	−2
2	0.31	0.43	0.97	0.025	0.010	−15
3	0.30	0.48	0.75	0.019	0.018	−26
E						
1	0.32	0.47	0.66	0.015	0.035	+43
2	0.34	0.46	0.61	0.019	0.030	+36
3	0.32	0.34	0.72	0.016	0.033	−21

* Tested in Normalized Condition (V-Notch)

Discussion of Impact Test Results

The "as-cast" low temperature impact properties were both inferior and more variable than those obtained on material in any of the heat treated conditions. This applies to both the low carbon and the medium carbon grades. Again, as would be anticipated, it is clear that a normalizing treatment is superior to a full annealing treatment, the degrees of improvement being on the average a reduction of 77-86 degrees F, in the transition temperature. The extent of this improvement will, however, depend upon the sectional thickness of the test material at the time of heat treatment, and it is to be expected that the difference between the properties of annealed and normalized material would be less in castings of heavier section. Nevertheless, the observation is relevant in that routine mechanical tests are usually carried out on test blocks similar to those employed for the purpose of this investigation.

The generally superior properties of heats B and D will be evident, and a number of factors probably contribute to this. Apart from unrecorded process differences, it will be noted that heats B and D have lower phosphorus and higher manganese values (see Table 2). The influence of phosphorus, however,

TABLE III—Heat Treatments Applied to Carbon Steels

Heat No.	Supplier's Usual Treatment	Alternative Treatment	Alternative Treatment
0.15 - 0.20% CARBON			
A	7 hrs. 1742°F A. C.	.....	.....
B	4 hrs. 1688°F A. C.	.....	.....
C	2 hrs. 1652°F A. C.	2 hrs. 1688° F. C. to 1022°F	2 hrs. 1688°F F. C. to 1022°F ¾ hr. 1652°F A. C.
0.30 - 0.35% CARBON			
D	4 hrs. 1688°F A. C.	.....	4 hrs. 1688°F A. C. ¼ hr. 1598°F W. Q. 1 hr. 932°F, A.C
E	2 hrs. 1697°F F. C. to 662°F	2 hrs. 1680°F A. C.	2 hrs. 1697°F . F. C. to 662°F ¼ hr. 1598°F A. C.

NOTE: Maximum section thickness being heat treated was 1½ inches.

is not clear since it was observed that steel A-2, which has a phosphorus content appreciably lower than A-1 and A-3, gives poorer impact figures which would suggest that phosphorus content is not a controlling factor. A more important factor is undoubtedly the manganese/carbon ratio, a high value favoring good impact figures. This observation has, in fact, been noted by other investigators in the field, but from the data available it cannot be concluded that the manganese/carbon ratio has in itself any intrinsic importance, but only that carbon raises and manganese lowers the transition temperature, each element having an effect more or less independent of the other.

Improved low temperature impact properties, as compared with those obtained in the normalized condition, are secured in the medium carbon grade of steel when this is either in the annealed and normalized condition, or in the annealed and quenched and tempered condition. Either of these treatments is effective in securing a 15 foot-pound (V-notch) temperature some 50 - 68 degrees F. lower than that obtainable on material in the normalized condition.

The progressive enhancement of the low temperature impact properties of carbon steels in going from the annealed condition, through the normalized condition, to the annealed and normalized condition, demonstrates the importance of grain refining heat treatment. The properties obtainable in material in the double normalized condition would in consequence be anticipated to show an even further improvement over those which have been obtained in the annealed and normalized condition.

This observation may be important in relation to establishing reasonable impact figures for future specification purposes. Presumably, these would be based on the assumption that the castings would be in the normalized condition. Consequently the measure of risk associated with failure to meet these figures is diminished when it is appreciated that any substandard material stands a good chance of being brought into a condition which would meet the specification requirements, by means of the application of an additional normalizing treatment.

TABLE IV—Summary of The Effect of Heat Treatment on Impact Properties

Heat No.	Treatment	Temperature (°F) of 15 ft. lb. (V-Notch)	Brinnel Hardness at 15 ft. lb. Temp.	Transition Temperature*
0.15 - 0.20% CARBON				
A	As Cast	+108	124	+189
	Normalized	—31	136	+48
B	As Cast	+57	150	+144
	Normalized	—80	122	+12
C	As Cast	+82	132	+190
	Annealed	+3	128	+100
	Normalized	—49	139	+77
	Annealed & Normalized	—72	140	+46
0.30 - 0.35% CARBON				
D	As Cast	+205	168	+242
	Normalized	—15	159	+25
	Quenched & Tempered	—49	230	—15
E	As Cast	+208	155	+244
	Annealed	+77	143.5	+162
	Normalized	+34	167	+118
	Annealed & Quenched	+14	159.5	+104

* Temperature at 50% Fibrous Fracture

Conclusions

The following conclusions, emerging as a result of this investigation, are based upon tests carried out on cast carbon steel test blocks, poured from 15 separate basic electric arc furnace heats, all finished with a ladle addition of 2-2½ pounds per ton of aluminum.

1. Normalized material gives superior low temperature impact properties, as compared with material in the fully annealed condition.
2. Annealed *and* normalized material gives superior low temperature impact properties, as compared with material in the single normalized condition.
3. The best low temperature impact properties are favored by maintaining a high manganese/carbon content ratio.
4. The "as-cast" low temperature impact properties of carbon steels are both inferior and more variable than those obtained on material in the heat treated condition.

CAUSES OF CRACKING IN HIGH-STRENGTH WELD METALS

*By A. J. Jacobs, R. P. Sopher and P. J. Rieppel**

Steel foundrymen should be most interested in this research because the results are directly applicable to cast steels at high temperatures. The test procedure is such that the center section of a test bar is made molten, then solidified and the bar tested at high temperatures. This is the same thing as applying stresses to steel castings shortly after they have solidified. Even though the information is directed towards welds, the findings are related to steel castings and the welding of steel castings. It is for these reasons that a summary of this research is included in this Journal—The Editor.

This article is a condensed summary of experimental work at Battelle Memorial Institute on the "Causes of Cracking in High-Strength Weld Metals" under the direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center. The information obtained is valuable in devising preventatives for cracking.

A literature survey made early in the investigation indicated that the most common form of cracking in high-strength weld metals is hot cracking, which occurs because of poor strength and ductility as the weld deposit is cooling down to about 1500 degrees F. A hot-tension test was used to evaluate the influence of composition on the properties of weld metals in this hot cracking range. The test involves the use of a cylindrical specimen placed in a 4½ turn induction coil. A 40 KW Lapel sparkgap frequency converter was used to heat the center section of the specimen to molten steel temperatures of the order of 3000 degrees F.

Specimens were heated to about 3000 degrees F., then cooled to a predetermined temperature in the range from 2700 to 1800 degrees F., at which temperature they were fractured. The relationship of hot strength and ductility, as obtained from the hot tension tests, to the actual cracking resistance of the steel was investigated by comparing the hot tension test data with the results of weld-metal cracking tests. The weld-metal cracking tests are discussed in more detail later in the report.

Tests on SAE 43XX - Type Steels

The compositions of the 13 SAE 43XX-type heats tested are shown in Table 1.

It was observed that misch metal affected the hot strength and ductility of 4340 steel, low in sulfur and phosphorus (Heat 7), more than the 4340 heats with increased amounts of sulfur or the 4320 heat which was also low in sulfur and phosphorus. A misch metal addition of 3 pounds per ton: (1) increases hot ductility above a temperature of about 2150 degrees F; (2) displaces the point of maximum ductility from a temperature of about 2000 degrees F to a temperature of about 2350 degrees F; and (3) increases hot strength above a temperature of about 2350 degrees F.

TABLE 1—Composition of SAE 43XX-Type Steels Studied in Hot-Tension Tests

Heat	C	Mn	Si	Ni	Cr	Mo	S	P
1	0.38	0.76	0.26	1.91	0.83	0.25	0.005	0.008
2	0.45	0.64	0.33	1.88	0.82	0.26	0.008	0.005
3	0.59	0.66	0.29	1.93	0.81	0.26	0.006	0.007
4	0.46	0.66	0.20	1.94	0.83	0.26	0.037	0.026
5	0.46	0.77	0.25	1.89	0.90	0.24	0.008	0.004
6	0.46	0.69	0.21	1.83	0.82	0.21	0.006	0.004
7([a])	0.45	0.72	0.07	1.92	0.88	0.24	0.006	0.007
11-A	0.40	0.76	0.22	1.84	0.83	0.24	0.015	0.012
11-B	As above, plus 3 pounds per ton of misch metal.							
12-A	0.35	0.77	0.23	1.84	0.81	0.23	0.036	0.011
12-B	As above, plus 3 pounds per ton of misch metal.							
1-H	0.20	0.67	0.30	2.07	0.91	0.39	0.008	0.010
3-H([b])	0.23	0.67	0.28	2.16	0.91	0.38	0.008	0.010

([a]) 3 pounds per ton of misch metal added.

([b]) 4 pounds per ton of misch metal added.

An increase in sulfur content was found to raise the temperature above which the hot strength and ductility are improved by a misch metal addition.

The beneficial effects of misch metal on hot strength and ductility were not so great as it was hoped they would be from experience in the castings industry. It has been postulated that hot ductility depends on the type of inclusion formed between misch metal and sulfur, which in turn is supposed to depend on the ratio of misch metal to sulfur. No relation has been established between hot ductility and the ratio of misch metal to sulfur in this study. Not only is the ratio of misch metal to sulfur widely different for heats in which misch metal was most effective, but the ratio also varies widely for other heats in which misch metal was not so effective. Thus, there is no apparent reason, at present, for the inconsistent behavior of misch metal.

The properties of 4340 heats, low in phosphorus and ranging from 0.008 to 0.036 percent sulfur were compared to show the effects of sulfur. Increasing the sulfur content from 0.015 to 0.036 percent causes a decrease in hot ductility, but has no significant effect on hot strength. It was found that a sulfur content of 0.008 percent results in unex-

* Battelle Memorial Institute, previously published in The American Welding Journal, Vol. 34, November 1955.

pectedly poor strengths and ductility. In other words, there seems to be an optimum sulfur content at which hot ductility is at a maximum. Perhaps there would be no advantage in reducing sulfur below this optimum to counteract hot cracking.

A comparison of the properties of two high-sulfur 4340 heats which differ in phosphorus content revealed a substantial increase in both hot strength and ductility accompanying a decrease in phosphorus from 0.026 to 0.011 percent. However, the heats also differ in carbon content. A comparison of low-sulfur and low-phosphorus heats indicates that a decrease in carbon from 0.50 to 0.20 percent tends to increase the hot strength and ductility. However, the improvement shown for a carbon reduction from 0.46 to 0.35 percent, the amount by which the heats used for phosphorus tests differ, was not so pronounced, particularly in ductility, as the improvement shown for the phosphorus reduction. Also, by decreasing the carbon content there is a general displacement of maximum ductility to a higher temperature.

It appears that the role of phosphorus as a causative agent in hot cracking is a real one. However, the exact role of phosphorus is unknown. There are several possibilities. For example, phosphorus may form eutectic or brittle constituents, such as nickel phosphide (Ni_2P, mp 2305 degrees F), iron phosphide (Fe_3P, mp 2010 degrees F), and a solid solution of phosphorus in iron plus iron phosphide (mp 1920 degrees F).

Test on Selected Steels

The composition of the five selected heats studied are shown in Table 2. These heats simulate high-strength weld metal deposited with a coated electrode.

Testing indicated that either a reduction in sulfur and phosphorus contents, or a misch metal addition of 2½ pounds per ton is capable of: (1) increasing hot ductility; and (2) raising the temperature of maximum ductility. It is believed that the maximum ductility may be related to the solidus temperature. The misch metal addition increases ductility above a temperature of about 2100 degrees F. It does not seem to be quite so effective in increasing ductility or raising the temperature of maximum ductility, as is a reduction in sulfur and phosphorus, which was the finding in tests on SAE 43XX-type steels. The heat simulating AISI 86B40 steel (5-J) appears to have excellent hot strength but poor ductility.

TABLE 2—Composition of Selected Steels Studied in Hot-Tension Tests

Heat	Chemical Composition, percent								
	C	Mn	Si	Ni	Cr	Mo	S	P	Other
8-A	0.14	0.88	0.55	1.77	1.00	1.00	0.007	0.006	0.19V
8-B	As above, plus 3 pounds per ton of misch metal.								
9-A	0.15	0.78	0.51	1.73	0.93	1.03	0.031	0.026	0.19V
9-B	As above, plus 3 pounds per ton of misch metal.								
5-J(a)	0.41	0.91	0.28	0.51	0.53	0.24	0.009	0.021	0.0015B

(a) 2½ pounds per ton of metal added; original sulfur content was 0.022%. Material was supplied by Metallurgy Department, Carnegie Institute of Technology.

Weld - Metal Cracking Tests

Two series of restrained weld tests were conducted to determine whether or not the results could be correlated with the results from hot tension tests. In the first series, 43XX-Type steel wire was deposited on base plate of the same composition. In the second series, (a) wire from the heats shown in Table 2 was deposited on low- and high-sulfur SAE 4340 steel base plate; (b) low-sulfur SAE 4340 wire, with and without misch metal, was deposited on high-sulfur SAE 4340 base plate; and (c) wire from an AISI 81B30 heat and a heat simulating AISI 86B40 was deposited on base plate of the same or nearly the same composition. Compositions of heats used in the restrained weld tests, which are not shown in Tables 1 and 2 are shown in Table 3.

The test involved straight weld beads joining two base plates which were restrained during welding and cooling. After welding, the plates were fractured in a bending press and analyzed for evidence of hot cracking.

All weld-metal cracking tests were conducted using the inert-gas consumable-electrode process. The wire for this process was prepared at Battelle by hot rolling and cold drawing the experimental heats to 3/32-inch. Frequent stress reliefs during the drawing operation were obtained by furnace anneal in vacuo and short time tempering in air by means of a welding generator. Chemical analyses of the wires showed no impurity pickup.

The results of the first series of tests on SAE 43XX steels indicates that: (1) increasing the carbon content of the steel in the range from 0.20 to 0.59 percent cause a moderate decrease in cracking resistance; (2) increasing the sulfur content in the range of 0.008 to 0.036 percent results in a pronounced decrease in cracking resistance; and (3) a misch metal addition to low-phosphorus SAE 43XX steels tends to increase cracking resistance.

The results of the second series of tests indicate that: (1) the weld deposits made with low-carbon, low-sulfur-phosphorus filler wire (Heats 8-A and 8-B) are more crack resistant than any other steel shown in Tables 2 or 3; however, if the sulfur and phosphorus contents are increased (Heats 9-A and 9-B) the cracking resistance becomes poor, regardless of the sulfur and phosphorus content of the base plate; (2) the 4340 welds, low in sulfur and phosphorus, exhibit poor cracking resistance when deposited on high-sulfur-phosphorus base plates, but are more crack resistant than welds made with the sulfur and phosphorus levels in the weld and plate reversed;

TABLE 3—Composition of Heats Studied in Restrained Weld-Cracking Tests[a]

Heat	Chemical Composition, per cent								
	C	Mn	Si	Ni	Cr	Mo	S	P	Other
4-A	0.40	0.70	0.24	1.85	0.80	0.25	0.034	0.032	.....
C	Commercial SAE 4340 steel containing 0.008 per cent sulfur.								
10-A	0.45	0.79	0.27	1.79	0.84	0.24	0.004	0.006	.....
10-B	As above, plus 3 pounds per ton of misch metal.								
404	0.36	0.85	0.49	0.20	0.43	0.11	0.015	0.076[b]	0.004B
Commercial AISI 81B30	0.33	0.99	0.28	0.32	0.49	0.15	0.013	0.017	0.002B

[a] See also Tables 1 and 2.
[b] Analysis of weld deposit made with Heat 404 wire in commercial 81B30 base plate.

(3) misch metal additions exhibited little effect on cracking resistance in these tests; and (4) the two boron steels have relatively poor resistance to hot cracking.

In view of the excellent performance of low-sulfur-phosphorus SAE 4340 filler wire when deposited in base plate of the same composition, the poor cracking resistance of this filler wire when deposited in high-sulfur-phosphorus SAE 4340 was quite surprising. Selected welds were analyzed for sulfur, and it was found that the welds contained about 0.020 instead of 0.005 percent sulfur. The poor cracking resistance was probably due to this extensive dilution.

It is best to employ a low-sulfur-phosphorus weld in all cases even though such a weld deposit is made on base metal of high sulfur-phosphorus contents. The reason for this is that there is less chance for sulfur to segregate in the center of the low-sulfur-phosphorus weld than in the center of the high-sulfur-phosphorus weld. The center of the weld is the last portion of the weld to solidify, therefore the low-melting sulfides would tend to segregate to this area. Hot cracking occurred most often in the center of the weld deposit.

Tests were made with two additional weld-metal cracking tests, namely the circular groove test and the circular patch test. An attempt was made to establish a correlation among the results of these tests, the restrained weld test and hot tension tests, thereby permitting the use of a less expensive test than the restrained weld test. However, both tests proved to be too insensitive to replace the restrained weld test.

The results of restrained weld-metal cracking tests are in satisfactory agreement with hot tension data. For example, decreasing sulfur and carbon contents were shown to increase hot ductility and resistance to cracking more than a misch metal addition. Yet the correlation has not been perfect because sulfur had a greater effect on cracking resistance but a smaller effect on hot ductility than did carbon. A reduction of phosphorus increased the hot strength and ductility substantially but its effect on cracking resistance was not definitely established.

RUSSIAN DEVELOPMENTS IN THE REPAIR WELDING OF STEEL CASTINGS BY THE CARBON DIOXIDE PROCESS

The Russian Central Institute for Engineering Technology (Welding Division) has been interested in the development of a process of fusion welding with a consumable electrode in an atmosphere of carbon dioxide, as well as in the design of a semiautomatic weld machine for welding casting defects with carbon dioxide as an inert protective gas.

The main feature of the semiautomatic welder developed at the Russian Institute consists of a mechanism for feeding the electrode wire through a flexible hose and holder, a carbon dioxide cylinder and gas drier, a generator and a power pack. The electrode wire, carbon dioxide, the water for cooling the holder and the welding current all pass through the flexible tube. The feed of the electrode wire is independent of the arc voltage.

The development of an electrode holder that provides for the simultaneous feeding of the electrode wire, gas, water, welding and operating current to the holder presented a serious problem, and more than 10 designs were tried before a satisfactory holder was developed.

The characteristics of the Russian Machine are:

(1) Fed from 3-phase line at 220 volts

(2) Welding current, d-c

(3) Maximum welding current - 500 amps

(4) Operating current for electrode feed motor - 36 volts

(5) Diameter of electrodes - $\frac{1}{16}$ - $\frac{5}{64}$-inch

(6) Rate of feed - 1½ to 80 feet per minute

(7) Length of flexible hose - 10 feet

(8) Carbon dioxide consumption - 35 cubic feet per hour

(9) Cooling water consumption - 8 gallons per hour

The rate of movement of the holder depends on the diameter of the nozzle and is limited by the danger of entraining air in the welding zone. The minimum distance between the nozzle and the surface of the casting is approximately ¾ - 1 inch. The mouth piece is embedded approximately ½-inch in the electrode holder in order to prevent short circuits, and the electrode projects about 1¼ to 1½ inches from the mouth piece.

The Russians experienced considerable difficulty with their equipment in obtaining good quality welds at the beginning and end of the bead. Also, porosity in the welds is not uncommon if care is not observed in the welding process. This weld porosity is attributed to the nitrogen in the air. The air can gain access as a result of leaks in the holder, high rates of movement of the holder during welding, excess gap between the tip of the holder and the surface of the casting, and a deficiency of carbon dioxide. Weld porosity is also attributed to voids in the casting just under the base of the weld repair area. During the welding the arc uncovers these voids and the gas entrapped in the voids escapes. This causes a deflection of the arc and raises a fountain of molten spray. The weld metal from such sites is reported to contain as much as 0.08 percent nitrogen.

For downhand welding the conditions for reversed polarity d-c are as follows:

(1) Welding current - 400-420 amps

(2) Arc volts - 30-32 volts

(3) Diameter of electrode - $\frac{5}{64}$-inch

(4) Rate of feed wire - 16 feet per minute

(5) Gas consumption - 35 cubic feet per hour

(6) Water consumption - 8 gallons per hour

(7) Electrode consumption - 15½-17½ pounds per hour

A series of experiments were conducted on the automatic and semiautomatic welding of steel castings employing carbon dioxide as an insert gas shield and using d-c with reversed polarity. A $\frac{5}{64}$-inch diameter filler rod was used. The composition of the standard filler rod was approximately 0.19 percent C, 0.91 percent Si and 0.88 Mn.

Several Russian steels were included in the investigation and their chemical compositions are shown in Table I.

TABLE I—Composition of The Steel Employed

Steel No.	C	Mn	Si	Cr	Ni	P	S
1	0.07	0.35	...	...	...	....	....
	0.12	0.50					
2	0.17	0.80	0.90	0.80	0.40	....	.....
	0.24	1.10	1.20	1.10	Max		
3	0.14	0.40	0.12	...	...	....	....
	0.22	0.65	0.30				
4	0.20	0.45	0.15	0.30	0.30	0.045	0.045
	0.25	0.70	0.30	Max	Max		
5	0.25	0.45	0.15	0.30	0.30	0.045	0.045
	0.35	0.70	0.30	Max	Max		
6	0.30	0.45	0.15	0.30	0.30	0.045	0.045
	0.40	0.70	0.30	Max	Max		

The composition of the deposited weld metal is affected by both an increase in current and voltage. The effect of increasing the current is shown in Table 2.

When the welding current was held at 400 amperes, increasing the voltage from 30 volts to 45 resulted in a continual lowering of the amount of manganese and silicon retained in the weld metal. It was also observed that as the voltage is increased the amount of oxygen and nitrogen in the weld metal increases, and when the arc voltage exceeded 42 volts porosity appeared in the weld metal. Increasing the voltage has only a slight effect on the tensile and yield strengths of the deposited weld metal but it has a marked decreasing effect on both the percent elongation and the percent reduction of area.

TABLE 2—Affect of Increasing Current on Composition of Weld

Metal		Current Amperes	Percent C	Percent Si	Percent Mn
Base Metal		.....	0.26	0.21	0.84
Filler Rod		.....	0.19	0.91	0.88
Weld	Rev	300	0.18	0.40	0.77
	Pol	400	0.21	0.37	0.79
Metal		500	0.22	0.29	0.72
	St	200	0.13	0.50	0.64
	Pol	300	0.14	0.43	0.67
		400	0.14	0.42	0.71

Arc Volts 28 - 34; Welding Rate 59-1/2 feet per hour.

A series of experiments were conducted to determine the effect of the composition of the filler electrode on the mechanical properties and composition of the weld metal. Standard welding procedures were followed, i.e., reversed polarity, current (400-420 amperes), rate of welding (78½ feet per hour) and arc volts 30-32.

Three experimental filler rods were developed and used to determine which analysis gave the best combination of mechanical properties, i.e., tensile, yield, percent elongation, percent reduction of area and impact values. The carbon content of all the filler electrodes was kept below 0.11 percent in order to reduce the tendency for the weld to hot crack. The silicon and manganese content of the filler rod was varied. The composition of the filler rod and the physical properties of the weld deposit are shown in Table 3.

The best combination of physical properties came from No. 2 rods which contained silicon in the range of 0.70-1.10 percent and manganese in the range of 0.90-1.30 percent. The high manganese and silicon filler rod deposited weld metal of a higher strength and hardness but had considerably lower ductility and impact values.

Some work was done on large diameter ($\frac{1}{16}$-inch) filler rod but the results were not satisfactory and as the diameter was increased the degree of penetration decreased. However, hot cracking was observed to decrease as the voltage was increased because of the improvement in weld penetration.

TABLE 3—Effect of Filler Rod Composition on Physical Properties of The Weld Metal

Rod No.	Filler Rod Percent Mn	Filler Rod Percent Si	Weld Metal Properties* Tensile Strength psi	Yield Strength psi	% Elongation	% Reduction of Area
1	1.61	1.72	100,000	77,000	16.0	30.5
2	0.97	0.78	81,000	58,500	27.0	59.9
3	0.87	0.56	79,000	60,000	17.6	34.9

* It is assumed that the physical properties of the weld metal are in the as welded condition.

Also, increasing the voltage was observed to cause a burning-out of the alloying elements as well as causing an increase in gas saturation. The best results were obtained by welding at a wire speed of less than 40 feet per hour on straight polarity. The reduction of the welding speed results in an increase in the ratio of the weld bead width to the weld depth. When the ratio of the weld width to the weld depth exceeded 1.3, hot cracks were absent.

The data reported justifies the following conclusions:

1. Production scale testing of the Russian semi-automatic machine confirmed the efficiency of semi-automatic welding of casting defects under a carbon dioxide shield.

2. The current, voltage, polarity and rate of welding have a pronounced effect on the chemical composition of the deposited weld metal.

3. When the voltage exceeds 36 volts the amount of gas in the weld metal is increased.

4. Hot cracking was eliminated in the first welding passes of multipass welds by reducing the welding rate to 40 feet per hour and by using straight polarity.

5. It is not necessary to clean or peen the welds between passes.

6. For welding steel castings in the downhand position, the following conditions were recommended: 400-420 amperes, reversed polarity 30-32 volts, 35 cubic feet of gas per hour.

7. The filler rod for welding castings produced in steel Numbers 3, 4 and 5 should not contain over 0.11 percent C. The following silicon and manganese contents are recommended: 0.8-1.10 percent Si; 1.00-1.30 percent Mn; 0.40 percent maximum P and 0.030 maximum S.

Very little has been published in this country on the repair welding of steel castings by the carbon-dioxide-shield consumable-electrode arc welding process. However, favorable comments on the process were mentioned by Dr. D. H. Rice and L. H. Lane of General Electric and Bernard Pizzino, Massillon Steel Castings at the 1955 T & O Conference.

The equipment developed in the United States consists essentially of a constant potential welder, a wire drive unit and a water cooled gun type electrode holder. The wire drive unit is geared so that when the wire speed is changed the welding amperage is automatically changed.

There is nothing startling new in the Russian investigation. The loss of elements in welding with mild steel electrodes by this process had been reported in the American Welding Journal. It is interesting to note that the explanation of weld porosity by American welding engineers is entirely different than the Russian theory. It is believed that the main source of porosity in carbon steel is the combination of carbon with oxides already in the base metal. Also, the carbon monoxide formed tends to become trapped by the freezing metal before the gas escapes.

Due to the automatic controls on the welding equipment developed in this country for this process a novice can make a reasonable quality weld.

Considerable work has been done in this country on the use of gas shields of carbon dioxide with argon, argon with oxygen, argon with helium and inert gases alone. No references were made in the Russian works to any other gas shields but the carbon dioxide shield.